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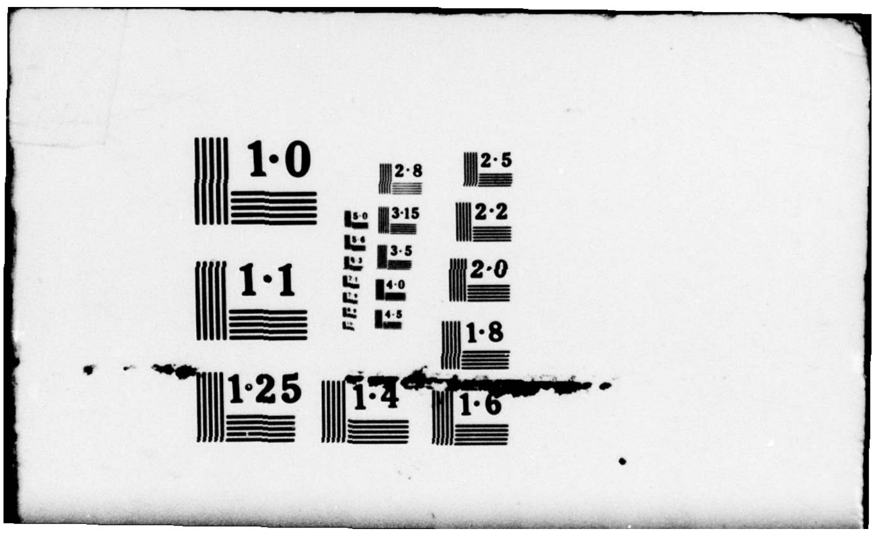
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LEVEL II

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FATIGUE CRACK NUCLEATION AND PROPAGATION IN ALPHA-BETA TITANIUM ALLOYS

by

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Final Report on AFOSR Grant No. AFOSR-75-2774.

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20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

Observations have been made of crack path behavior of α-β Titanium alloys on the surface and in interiors. The fracture surface for an α(Ti-0.4Mn) a β(Ti-10.2Mn), and an α-β alloy (Ti-5.2Mn). have been examined, TEM examination of α and β alloys carried out and a study of slip line reversal during load reversal was made.

Surface cracking occurs along α-β interfaces and these cracks are joined by link-up. In the interior the crack proceeds along the lines of easiest slip

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* alpha ϕ beta

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with the result that cracks do not necessarily follow interfaces. The concentration of the crack path in α or along α-β interfaces is attributed to higher strain in these regions.

Striations are found to be more prominent in β than in α and this is attributed to the higher strain hardening of the α phase. Hardening in α was attributed to an increase in dislocation density and interaction between twins and dislocations. It was not possible to ascertain whether softening in β was due to an increase in dislocation density, a decrease in density of the structure giving rise to streaking on diffraction patterns or both.

Slip line reversal during reverse straining was found to occur preferentially at grain boundaries. During reverse straining at strain below the forward strain new slip lines appeared indicating that complete slip reversal did not take place. Slip height levels were found to be higher in the new slip bands than in the slip bands which operated in both tension and compression.

* alpha
ø beta

I. FINAL TECHNICAL REPORT

A. Introduction

This investigation has been concerned with fatigue crack nucleation and the early stages of propagation in Ti-Mn alloys with equiaxed (E) and Widmanstätten plus grain boundary (W+GB)α structures. The range of Mn contents, from 0.42 to 15.4 Mn (compositions in weight percent) permits α or β to be matrix when two phases are present, and the Mn content is sufficient to permit the β to be retained on quenching from 700°C.

The Ti-Mn system was used as a model system to permit the role of morphology and volume fraction of phases in fatigue crack initiation and early propagation to be ascertained, while keeping the intrinsic nature of the two phases constant by keeping their compositions constant.

During this past year, November 1, 1977 to October 31, 1978, our program was completed. This report will outline the work carried out during this period, present the highlights of the results and will discuss their significance.

B. Experimental Results and Discussion

During this past year we have examined crack path behavior on the surface of fatigue specimens and in the interior; we have studied the fracture surfaces of fatigued specimens; made transmission electron microscope observations and have studied the behavior of slip reversal in specific grains of a Ti-15.4Mn alloy.

1. Fatigue Path Observation

a. Equiaxed (E) Structures

Grain boundary cracking was observed in single phase, beta alloys, but this was not the predominant mode of cracking at high strains. This is attributed to the fact that slip in beta is coarse, rather than dispersed, and as a result large slip offsets are produced which are conducive to crack initiation.

In the α-β alloys, cracking was shown to be mostly associated with α, either α-β interfaces or α itself. This can be seen from the data of Table I, which lists the volume fraction of phases along the crack. It can be seen that the sum of the volume fraction of α along the crack plus interface α exceeds the volume fraction of α in the alloy. For this purpose we are considering travel along the interface as travel in α. In all but one case in Table I this excess is essentially equal to the volume fractions of interface α. The crack is thus seen to prefer to travel in or along α. In work that we have carried out, using finite element analysis (1), it has been shown by both calculation and experiment that the average strain in α is greater than the average strain in β, within the strains used in our fatigue tests. Further the calculations have shown that the strain in β is higher at the α-β interface than in regions in β away from the interface. Also the average strains in α are likely to be higher within the particle than they are at the interface. Thus the crack would appear to follow those regions of the microstructure where strains are highest.

b. Widmanstätten + Grain Boundary (W+BG)α Structures

At the surface there is considerable preference for cracks to form along α-β interfaces. Slip initiates at α-β interfaces and,

Table I. Crack Path Measurements in Alloy 5, Ti-5.2Mn, Equiaxed Alpha Structure

Interior or Surface Crack	Particle Size μm	Volume Percent of Alpha %	$\Delta\epsilon_T/2$, %	Volume Percent Phases Along Crack		
				α Phase	β Phase	α-β Interface
Interior	4.3	56	0.8	59	26	15
Surface	4.3	56	0.8	59	29	[illegible]
Surface	4.3	56	1.5	56	23	21
Surface	10.4	56	0.8	54	20	27
Surface	10.4	56	1.5	37	31	32

consequently, one would expect cracking to commence in these areas. Once cracking at Wα-β or GBα-β interfaces occurred, link-up between the cracked regions would take place because deformation would be concentrated at the crack tips. It would, therefore, be inappropriate to conclude that crack propagation preferred α-β interfaces. Crack nucleation takes place much more readily in W+GBα structures than in Eα structures, because they provide long slip paths which are conducive to crack nucleation. It should be noted that the long surface cracks in the W + GBα structures do not penetrate as far into the specimen as surface cracks do in Eα structures before unstable fracture begins. These long surface cracks undoubtedly lead to the shorter fatigue lives observed for the W + GBα structures. Thus, one cannot conclude, from observations that crack growth in W + GBα are slower than in Eα structures, that fatigue life will be longer in such structures. Such measurements have been made in specimens where the crack length has been predetermined by the specimen design. In real structures such limits are not automatically established.

Crack growth in the interior of the specimen, that is in the stage two region, appears to take the path of easiest slip. Such a path cuts across colonies of Widmanstätten α and will, on occasion, pass along an α-β interface. This seeking of the easiest slip path results in a large fluctuation of the crack around the general crack path. The sharp changes in crack path must contribute to the slower rate of grack growth for the W + GBα structures. Slow growth of cracks in W + GBα structures is also enhanced by multiple cracking which tends to reduce the stress intensity at the crack tip.

It was found that GBα was thicker than Wα when the α was precipitated by reheating into the α-β field after quenching from the β field. The two types of α were found to be equal in size when the W + GBα structures were formed on cooling directly from the β to the α-β field. The coarser grain boundary α, where slip could take place more readily, produced cracking more readily than the Wα.

2. SEM Examination of Fracture Surface

Fractographic examination was carried out on three alloys, the α alloy #3, Ti-0.4Mn, the β alloy #7, Ti-10.2Mn, and an α-β alloy, alloy #5, Ti-5.2 Mn. The observations were carried out primarily in the "thumbnail" region of the fracture surface. In general striations were not well formed in α but were clearly formed in β. This behavior was also evident in alloy #5, where the difference in striation behavior permitted distinctions to be made between equiaxed α and the β-matrix. In W + GBα structures the shape of the α as well as the reduced tendency for striation formation delineated the α.

The reduced tendency for striation formation in α would seem to be related to the larger-strain hardening which takes place in α. Consequently, the enhanced striation formation in β would be related to its lower strain hardening capacity. There are some interesting implications to this behavior. If one examines the true stress true strain behavior of the α and β alloys, one notes that the strain hardening in the early parts of the stress-strain curves is higher in α but that at higher strains β strain hardens much more than does α. Thus, it appears that stress-strain behavior at the low strains, used in the cyclic tests, governs fatigue behavior. Also, it must be concluded that the intensifi-

cation of strain at the crack tip does not extend the stress-strain behavior into the high strain hardening region of the β except near the end of the striation formation region, where tensile failure begins.

A series of observations were made at increasing distances from the surface of a specimen of alloy #5, Ti-5.2Mn with E structure. Striation spacing and depth were found to increase with increasing distance from the surface, which corresponded to increasing ΔK. Secondary cracks were found to occur at the bottom of striations and such occurrence was independent of whether the structure was E or W+ GBα. At still further distances from the surface voids were found mixed with striations, an observation which indicated that tensile failure had begun. Although this behavior indicates that strain levels had reached the high strain hardening levels of the β stress-strain curve, the onset of void formation prevents any considerable benefit from occurring, since failure occurs in a relatively few additional cycles.

3. TEM Observations

TEM observations were carried out to obtain some insight into cyclic hardening behavior. The α alloy, after annealing for 200 hrs at 700°C, had only revealed isolated dislocations. The alloy would then be expected to harden cyclically, and this behavior was observed. The hardening was consistent with the results of Stevenson and Breedis (2) at high plastic strains.

The upswing noted in the average peak stress vs. number of cycles at later stages of cycling was correlated with the occurrence of a large volume of twins (2). Extensive twinning, especially at higher strains, was also noted in the present work. This twinning

could, conceivably, be related to the continued hardening as a result of interaction between slip and twinning.

The cyclic softening observed in the β alloys could arise from an increase in mobile dislocation density as first suggested by Theodorski and Koss (3) and subscribed to by Chakrabortty et al (4). However, in view of the streaking which has been observed in the diffraction patterns of β, metallurgical instability could not be ruled out. Attempts were made to determine whether cyclic deformation changes decreased the intensity of streaking. However, it was found that the intensity of streaking varied from grain to grain, and spontaneous transformation of β to martensite also interfered with the results. It was not possible to reach any conclusions regarding the role of metallurgical instability on softening.

Softening was found in α-β alloys, where volume fraction of phases would suggest that hardening would occur. TEM examination indicated that softening occurred because dislocations, initially uniformly distributed were gathered into patches. As a result long dislocation free paths were established.

4. Slip Reversal in β Ti-15.4Mn Alloy

If slip during cyclic straining could be completely reversed, crack initiation would not take place. It was decided to examine slip reversal in β, because slip was much coarser in β than in α, and, therefore, easier to observe. A series of eighty grains were examined after a forward tensile strain of 0.2%, T_1, a reverse compressive strain of 0.1%, C_1, and a reverse compressive strain of 0.2%, C_2. During T_1, slip was found to commence at grain boundaries and proceed inward. During C_1, slip reversal began at grain boundaries and proceeded into the grains. During C_1, slip on secondary slip systems was found to develop at a small fraction of

the grain diameter from the grain boundary. This secondary slip occurred during the Bauschinger region of reverse deformation, thus insuring that complete slip reversal did not take place.

Interference pattern measurements revealed that the difference in slip heights across the secondary systems were greater than the difference in slip height across the original slip systems after the C_2 strain. This behavior indicates that prior slip in the forward direction limits the extent of slip which can be achieved in the reverse direction.

It is of interest to consider what processes cause slip reversal to begin at the grain boundaries. Prior work (5, 6) has shown that the stress at the grain boundary is higher than the stress at the grain interior during plastic deformation. The stresses which occur at the boundary also effectively raise the apparent elastic modulus (7) because these stresses oppose the deformation which would accompany the applied stress. The net result is a stress-strain curve which is different for the grain boundary and the grain interior, Figure 1.

On unloading after an initial strain, the grain boundary region, because of its higher apparent modulus, unloads more rapidly than the grain interior. At zero total strain (see the center average curve) the grain boundary would be in compression and the grain interior in tension. When compressive stress is applied, the grain boundary would yield first,and,thus,slip would disappear first there. The fact that slip disappears first at grain boundaries would also contribute to the small extent of grain boundary cracking noted earlier.

C. General Comments

Our fatigue results together with our finite element analysis (1) lead to some rather interesting observations. The fatigue limit occurs

at stresses considerably below the yield strength. It has been concluded that in order for fatigue damage to occur some plastic deformation must occur. The finite element analysis has indicated that at stresses considerably below the yield stress of the alloy plastic deformation is occurring as a result of stress inhomogeneities at α-β interfaces. In single phase material at the low stresses involved, near the fatigue limit, grain boundary regions would be expected to yield first and hence to be the sources of fatigue initiated cracks.

The larger is the difference in yield stress between α and the β matrix. The lower is the ratio,between applied stress at which interface deformation takes place and the yield strength, likely to be. Thus, when higher yield strength α-β alloys are used one would expect lower fatigue limits relative to the yield strength. It would also be anticipated that for coarse α-β structures creep at room temperature would be encountered at lower fractions of the yield strength, the larger the difference between the yield strengths of α and β.

In the Ti-6Al-4V alloy the α is strengthened and consequently the ratio of yield strengths of the matrix to that of α probably does not exceed 1.6 to 1. even in the aged material. In our alloys the ratio of β yield strength to that of α is approximately 3 to 1. These observations will become more noticeable, industrially, as the yield strengths used rise.

D. Publications

Two papers have been accepted for publication:

1. Y. Saleh and H. Margolin: "Bauschinger Effect During Cyclic Straining of Two Ductile Phase Alloys," accepted for publication in Acta Met.

2. H. Margolin, F. Hazaveh and Y. Yaguchi: "Grain Boundary Contributions to the Bauschinger Effect," accepted for publication in Scripta Met.

Three other publications, previously reported, resulted from work on this grant.

E. Presentations

The following presentations were made at the AIME meeting October 15-19, 1978 in Denver, Colorado.

Y. Saleh and H. Margolin: "The Effect of Microstructure and Volume Fraction of Phases on Low Cycle Fatigue Behavior of α-β Ti-Mn Alloys."

H. Margolin: "The Role of Microstructure and Stress Gradients in Fatigue Crack Initiation of α-β Titanium Alloys."

II. PERSONNEL INVOLVED

The personnel involved in this study are Professor Harold Margolin and Yousef Saleh, who has completed his doctoral program and is now working in industry. Fakhreddin Hazareh has joined the program during the past year. He has worked on the study of slip line reversal during reverse loading.

III. COUPLING

Professors Donald Koss and Harold Margolin have discussed their respective research programs and Professor Margolin has sent material to Professor E. Starke for inclusion in his ASM review article on fatigue.

1. S. Ankem and H. Margolin: "Finite Element Method Calculations of Stress-Strain Behavior on Alpha-Beta Ti-Mn Alloys," presented at TMS-AIME Fall Meeting, St. Louis, MO, October 15-19, 1978.

2. R. Stevenson and J.F. Breedis: Acta Met, 1975, Vol. 23, p. 1419.

3. G. Theodorski and D. Koss: Met. Trans. 1976, Vol. 7A. p. 1243.

4. S.B. Chakrabortty, T.K. Mukhopodhyay and E.A. Starke, Jr.: ONR Tech. Rept. on Contract N00014-75-C-0349, NR 031-750.

5. Yii-der Chuang and H. Margolin: Met Trans. 1973, Vol. 4, p. 1905.

6. T.D. Lee and H. Martolin: Met Trans. A. 1977, Vol. 8A, pó 158.

7. T.D. Lee and H. Margolin: Scripta Met. 1977, Vol. 11, p. 713.

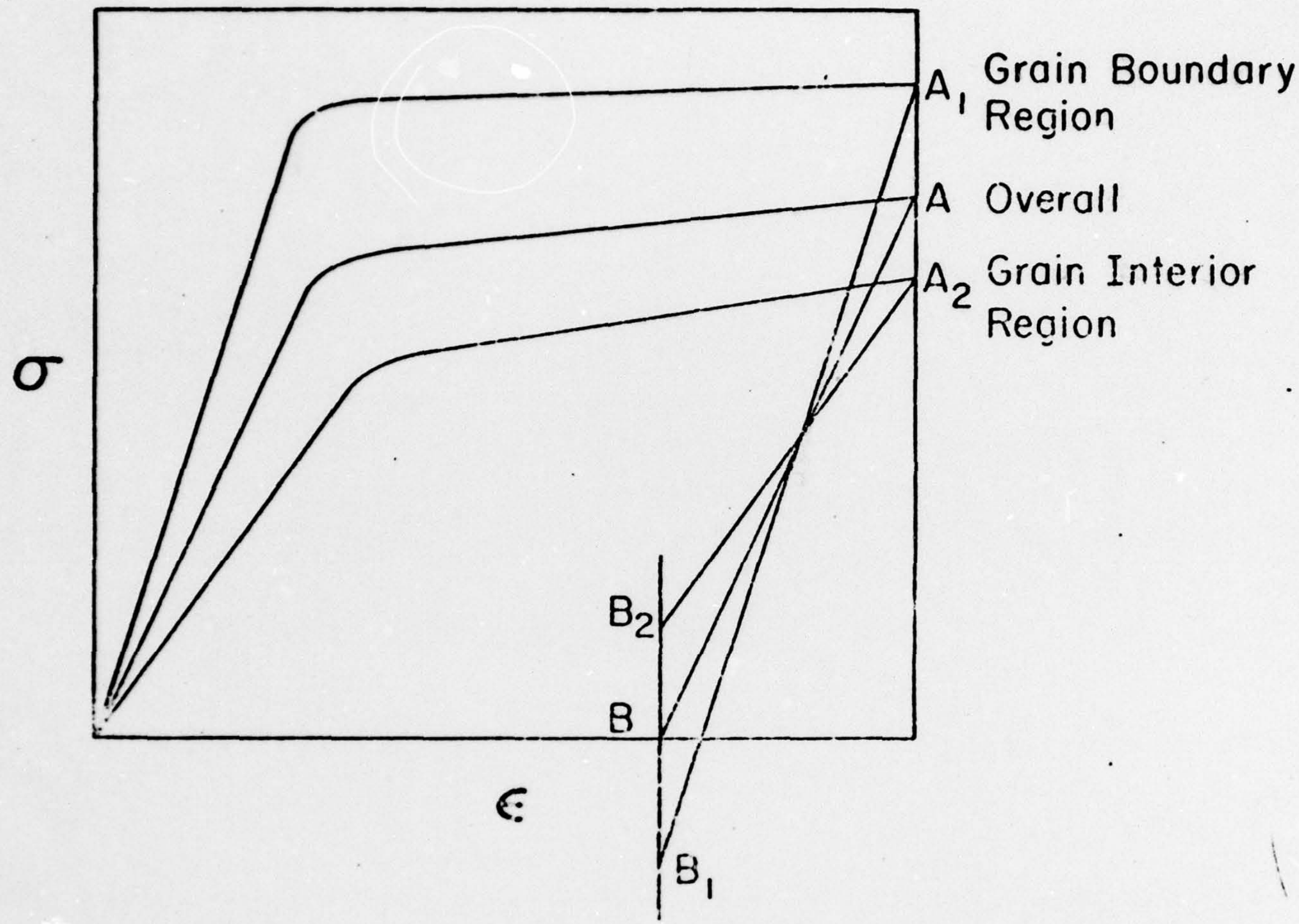

FIGURE 1. Stress Strain Curves of Grain Boundary and Grain Interior Regions and the Overall Stress Strain Curve. Also shown is the stress distribution on unloading to zero overall stress.